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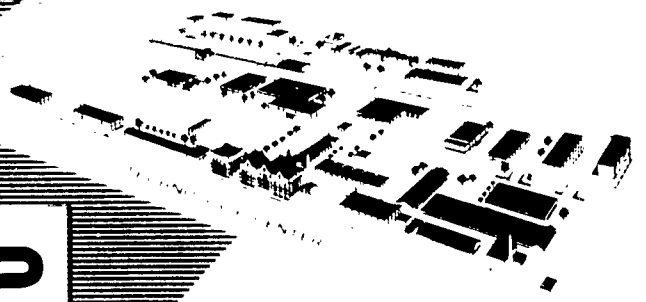
ARF 2185-12
(Final Report)

ARMOUR RESEARCH FOUNDATION OF ILLINOIS INSTITUTE OF TECHNOLOGY



ASTIA
APR 15 1958

WELDING OF ALUMINIZED STEEL

Contract No. NObs-77183

Index No. NS 021-200

Chief
Bureau of Ships
Department of the Navy
Washington 25, D. C.

Attention: Code 637

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ARMOUR RESEARCH FOUNDATION

of

Illinois Institute of Technology
Technology Center
Chicago 16, Illinois

ARF Project No. 2185

Contract No. NObs-77183

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ARF 2185-12
(Final Report)

July 1, 1959 to June 30, 1960

for

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Bureau of Ships
Department of the Navy
Washington 25, D. C.

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July 12, 1960

WELDING OF ALUMINIZED STEEL

ABSTRACT

The manual metallic arc welding of aluminized steel was investigated to determine the suitability of aluminized steel for ship construction from the standpoint of welding. Various electrodes were investigated, but only the low-hydrogen types proved capable of depositing porosity-free welds.

The all-weld metal tensile, transverse tensile, longitudinal and transverse fillet weld shear, notch toughness, and explosive impact properties of mild steel (Grade M, Mil-S-16113B) plate coated with 1100 aluminum, and welded with low-hydrogen electrodes, were determined. The results compared unfavorably with similar properties determined for welds made in uncoated plate. The explosive impact properties were particularly unfavorable, leading to the conclusion that mild steel plate coated with 1100 aluminum welded with low-hydrogen electrodes is not suitable for naval ship construction.

Unsuccessful attempts were made to flux the aluminum from the faying surfaces of the weld to permit unalloyed weld metal. Both preplaced flux and flux dip-coated onto electrodes were tried.

A discriminating explosion test applicable to fillet welds was developed which promises to be very useful in investigations of the type covered herein. Additional work needs to be done to fully evaluate the test variables.

Preliminary investigation of Grade M plate coated with aluminum alloy K726A (2 1/2% silicon) and welded with low-hydrogen electrodes, was more encouraging than that of the Grade M plate coated with 1100 aluminum alloy. Specifically, the welds had considerably more ductility and higher strength, and were free of the extremely hard aluminum-rich compounds Fe_2Al_5 and $FeAl_3$ found in the welds of the 1100 aluminum-coated steel. Additional work is required to confirm these results, which would then open the possibility of making successful, tough welds in aluminized plate by modifying the composition of the aluminum alloy coating.

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WELDING OF ALUMINIZED STEEL

I. INTRODUCTION

Corrosion in naval vessels has prompted the Bureau of Ships to search for economical corrosion-resistant materials suitable for ship construction. In the past, galvanized steel has been used for this purpose, but fillet weld cracking caused by the penetration of liquid zinc into the root of the welds in the presence of stress has created considerable concern with regard to the battle performance of ships fabricated using galvanized steel.

The problem of fillet weld cracking of galvanized steel was investigated at the Foundation on a prior program, and the mechanism of cracking was positively identified. Several means of eliminating the cracking were presented, but none of them have been shown to offer, practically, the assurance of freedom from cracking needed in naval vessels.

The Bureau of Ships then became interested in other corrosion-resistant materials, one of which is aluminized steel. A preliminary investigation by Baysinger of Kaiser Steel indicated that welds in aluminized (K726A) steel may be crack sensitive unless special welding techniques are used to reduce restraint. This program was initiated at the Foundation to investigate the weldability of aluminized steel and to determine the suitability of this material for naval vessels subjected to explosive shock loadings.

II. MATERIALS

A. Base Metals

Due to the commercial unavailability of aluminized steel in plate thicknesses, hot rolled plate steel was purchased and aluminized at the Foundation. Although both Grade M and Grade HT steel, Mil-S-16113B, were ordered, only the Grade M type plate was used on this program. All the plate used in this study was 1/2 inch thick.

The mill analyses (Inland Steel) for the material were:

C	Mn	P	S
.25	.48	.010	.023
.27	.43	.010	.030
.23	.41	.010	.030

A check analysis on one heat of this material showed:

C	Mn	P	S	Si	Al
.23	.48	.008	.026	.02	.02

The Grade M plate was aluminized with both 1100 and K726A aluminum alloys. The mill analyses for these coatings were:

Alloy	Al	Si	Fe	Zn	Cu	Cr	Ti
1100	99.81	.10	.07	.02			
K726A	Remainder	2.53	.44	.01	.01	.12	.04

Because of the batch dipping method of coating described in a subsequent section, the coating thicknesses varied from sample to sample and from one end to another even on the same sample. The coatings varied from about 1-3 to 6-8 thousandths of an inch from top to bottom*, except on one end of each sample where a much heavier coating formed when the samples were hung on the rack to cool after dipping. No great effort was expended to get uniformity of coating thickness on an individual sample since, with the variation from thin to thick, it was possible to obtain the effects of varying aluminum content on each welded specimen. It was reasoned that the extensive effort required for development of uniform coating thicknesses could be postponed until the investigation proved them necessary.

B. Aluminizing Procedure

The steel plate samples were vapor degreased, then pickled in 20% HCl at 160°F to remove mill scale, and water rinsed. They were subsequently immersed in an aqueous solution of sodium silicon fluoride, zirconium tetrachloride, and potassium chloride composed as follows:

10 gm Na_2SiF_6
2 gm $ZnCl_4$
0.5 gm KCl
1 liter H_2O

After immersion, the plates were allowed to dry in moving air which left a salt crystal residue on the plate surfaces, thus providing a mild flux for the aluminizing operation. Since this crystalline flux was ineffective in

* Corresponding to the position of the samples when cooling after dipping.

preventing reoxidation of the steel, it was necessary to dip the plates within 20 minutes of the flux drying operation.

When the plate samples were absolutely dry, they were dipped in molten aluminum alloy of the desired analysis which had been degassed with chlorine. The recommended bath temperature according to one of the major aluminum suppliers is 1260°-1270°F; but because of the relative heat capacities of the low volume bath and the steel plates, it was necessary to superheat the bath of 1100 aluminum to 1400°F before dipping each plate. When this temperature was reached, the surface of the bath was carefully skimmed to remove the aluminum oxide film; then the plates were slowly lowered into the molten aluminum alloy. This caused the bath to drop to 1200°-1250°F, depending on the sample size. The plates were allowed to remain in the bath until the bath temperature increased to 1350°F which required 3-6 minutes, again depending on the sample size. At this point, the bath surface was again carefully skimmed of oxide and the plates were quickly withdrawn, shaken to remove the excess aluminum, and hung on a rack to cool with the flat sides of the plates in the vertical position. Just before the aluminum solidified on the steel, the lower part of the plates was carefully wiped with a piece of steel strip to remove as much of the excess aluminum as possible.

Somewhat lower temperatures were used for the K726A aluminum alloy coating because of the lower melting point of this alloy.

C. Welding Electrodes

All the welding done on this program was performed with coated electrodes of the following sizes and types:

Electrode Types	Size, in.	Supplier
E 6010	5/32	A
	1/8	B
	5/32	B
	3/16	B
	1/4	B
E 6010 iron powder	5/32	A
E 6012	5/32	A
E 6015	5/32	C
	1/4	C
E 6016	5/32	A
E 7016	5/32	C
	1/4	C
	5/32	B
	3/16	B
	5/32	D
	3/16	D
E 7018	5/32	A
E 6020	5/32	A
E 6024	5/32	A
E 6027	5/32	A
Raco Fer-Al	1/8	C
	5/32	C
	3/16	C

III. EXPERIMENTAL PROGRAM

A. Development of the Initial Problem

At the beginning of this program, there was very little published information concerning the welding of aluminized steel. Therefore, it was necessary to run some preliminary experiments to define the problem.

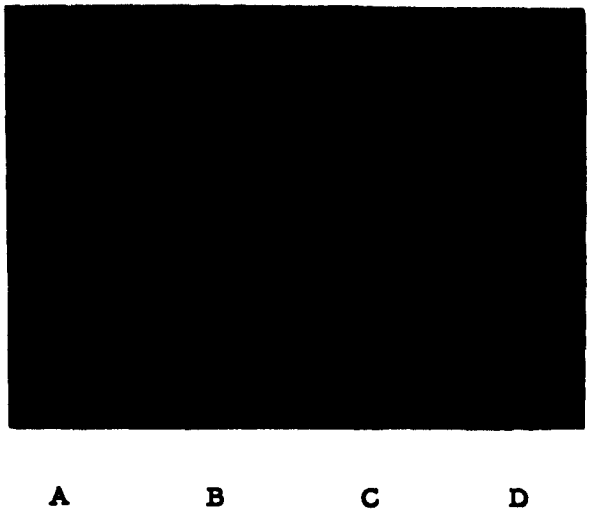
The first experiments were a series of beads deposited on a plate coated with 1100 aluminum alloy using 5/ 32 inch diameter E 6010, E 6012, E 7016, E 7016 iron powder, E 6020, E 6024, and E 6027 electrodes. All the beads exhibited gross surface porosity with the exception of the beads deposited with the low-hydrogen electrodes, E 7016 and E 7016 iron powder (see Figure 1). The porosity was worse for the beads deposited with E 6010 electrodes, and was somewhat less, though excessive, for the E 6012, E 6020, E 6024, and E 6027 beads, in order of decreasing porosity. Similar experiments were run using various joint designs including fillet, square butt, and double-vee beveled joints, with the same results. No porosity was observed for any of the welds made with E 7016 or E 7016 iron powder electrodes.

At this stage, the investigation was divided into two parts: one part was the investigation of ways and means of eliminating porosity in welds made with cellulose-coated electrodes, and the other was the further investigation of aluminized steel weldability using low-hydrogen electrodes. To this end and in order to reduce the number of specimens required, E 6010 electrodes were selected to represent the cellulose-coated electrodes, and E 7016 electrodes were selected to represent the low-hydrogen types.

B. Porosity With Cellulose-Coated Electrodes

The results of the bead-on-plate experiments indicated a general correlation between the amount of cellulose in the electrode coatings and the amount of porosity produced. This, together with the results of the experiments with low-hydrogen electrodes, indicated that the porosity was caused by the evolution of hydrogen from the weld metal during solidification.

Since the solubility of hydrogen in molten steel is increased by the addition of aluminum and since the solubility decreases sharply upon solidification, it was not felt that variations in welding procedure to decrease the



A B C D

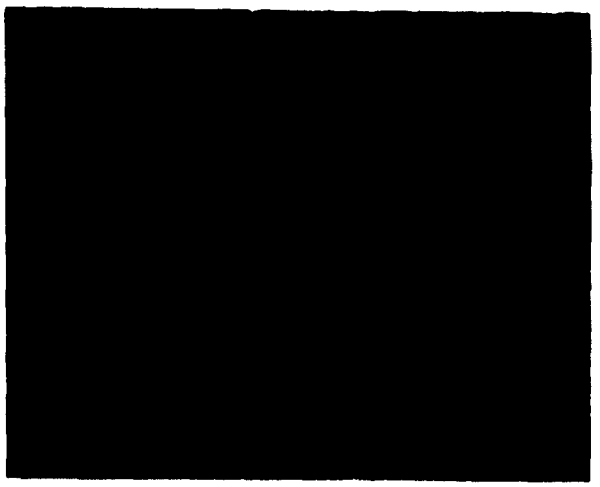
Electrode Type

A = E 6010

B = E 6012

C = E 6020

D = E 6024



E F G H

E = E 6010

F = E 7016

G = E 7016 Iron Powder

H = E 6024

Figure 1 - Appearance of Surface Porosity Obtained With Cellulose-Coated Electrodes Compared To Low-Hydrogen Electrodes (Material: Grade M Coated With 1100 Aluminum Alloy)

cooling rate of the weld would be very effective in eliminating the porosity. However, experiments were performed to check this: bead-on-plate welds were made using a weaving technique with preheats up to 500°F. Gross porosity still existed. It was observed that the evolution of hydrogen took place over a wide temperature range indicating a wide variation of aluminum content throughout the weld metal. Metallographic examination verified this. (See Discussion of Work and Results). However, the porosity was never positively observed as originating in the aluminum or aluminum-rich phases.

It was next reasoned that the manner of weld deposition--together with incomplete mixing of the weld metal, coating, and base metal--produced a weld with a higher iron content in the surface regions than in the root regions. Since the aluminum lowers the melting point of the steel, the top of the weld can freeze first and the porosity can originate in shrinkage cavities being fed by hydrogen evolved from the lower melting aluminum-alloyed weld metal. This mechanism does occur in some instances, as shown in Figure 2 which is a cross section of a large-porosity cavity at the end of a weld. The top half of the "bubble" was sawed off and analyzed for aluminum, which was found to be 0.01%. The average aluminum content for the weld was 0.81%.

These facts led to the conclusion that elimination of porosity in welds deposited with cellulose-coated electrodes would require either the elimination of the aluminum from the faying surfaces, or else the reduction of weld metal aluminum content to a much lower level with the aluminum more uniformly distributed.

Several experiments were performed in an effort to flux the aluminum from the faying surfaces. The basis for these experiments was the possibility of converting the aluminum to an oxide which would then become part of the weld slag.

Fluxes of two basic compositions were used. One flux had as its essential ingredient Fe_2O_3, which was intended to be reduced by the aluminum to give the thermit reaction producing iron plus aluminum oxide. The other flux had as its essential ingredient $CaCO_3$ which was supposed to react with aluminum to produce calcium aluminate plus carbon dioxide which would go off as gas. The iron oxide base fluxes were ineffective in eliminating the

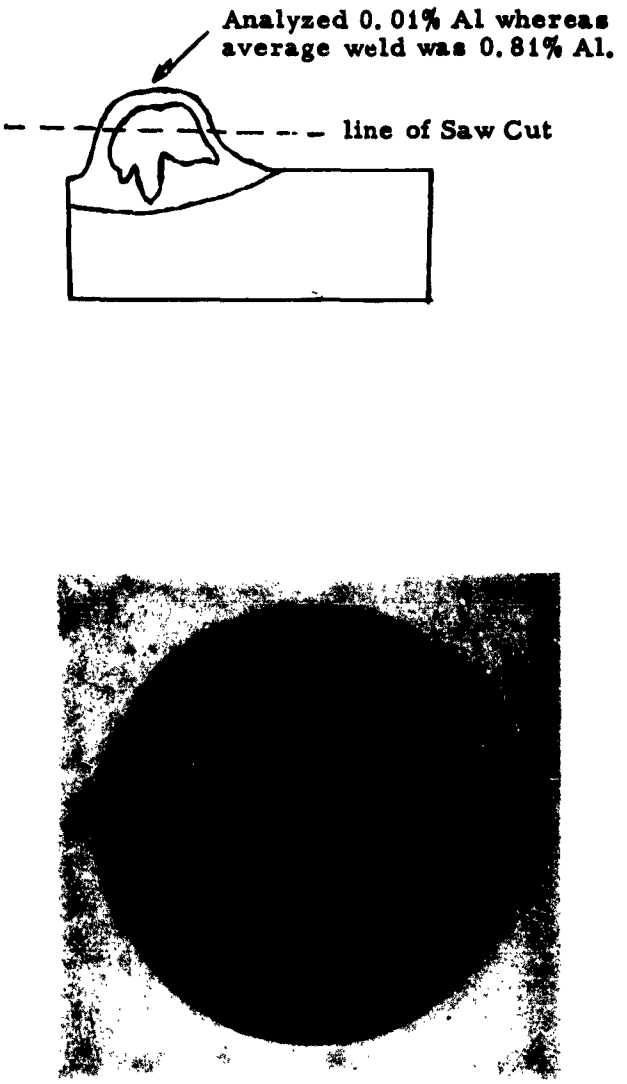


Figure 2 - Cross Section of Weld Showing Large Porosity Bubble.

porosity, but beads almost free of surface porosity were obtained with a calcium carbonate base flux composed of 20% $CaCO_3$, 20% CaF_2 and 6.0% Na_2SiO_3 (44 Baumé) by weight. However, when this flux was applied to joints such as tee and square butt types, gross porosity reappeared apparently because of the aluminum entrapped in the closed part of the joints.

Other experiments were performed in an effort to reduce the aluminum content to a point where it would be ineffective in promoting porosity. The principle behind these experiments was to dilute the weld metal either with base metal, filler metal, or both. Electrodes from 1/8 in. to 1/4 in. and at varying arc lengths, amperages, and weld speeds were tried without success.

At this point, in view of the porosity-free welds obtained with low-hydrogen electrodes, all work with cellulose-coated electrodes was discontinued.

C. Weldability With Low-Hydrogen Electrodes

In view of the porosity-free bead-on-plate welds obtained early in the program using low-hydrogen electrodes, further evaluation of the weldability of aluminized steel using low-hydrogen electrodes was undertaken. The following factors were investigated:

a. Effect of moisture content of the electrode coatings on porosity and weld ductility.

b. Performance of aluminized steel when subjected to the cruciform test.

c. Mechanical properties of welds in aluminized plate including all-weld-metal tensile properties, transverse tensile properties, transverse bend properties, longitudinal and transverse shear properties, and notch toughness.

d. Performance of fillet welds in aluminized steel plate subjected to explosive loadings.

1. Effect of Moisture Content of Electrode Coatings on Porosity and Weld Ductility

The effect of moisture content of the "low-hydrogen" electrode coatings on porosity and weld ductility was determined by aluminizing previously

machined double-beveled plates, and making butt welds with these specimens using 3/16 in. diameter low-hydrogen electrodes containing 1.2%, 0.5%, and 0.17% moisture (see Figure 3).

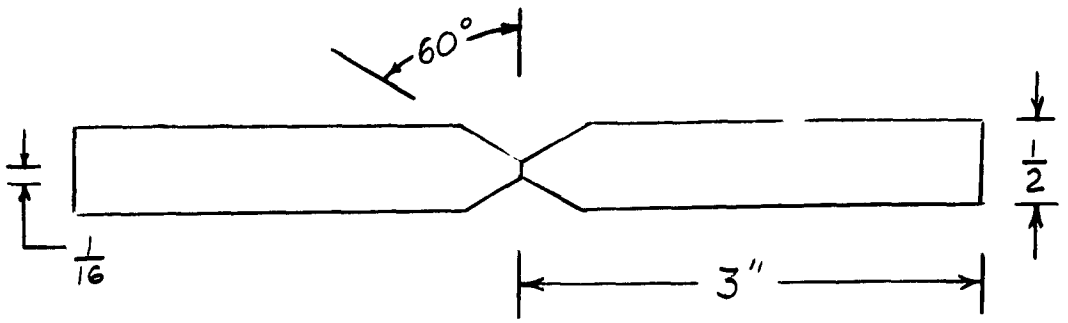


Figure 3 - Joint Design Used for Butt Welding Specimens.

The electrodes were humidified by exposure to the atmosphere for several weeks; then the 1.2% moisture content electrodes were used as-humidified, the 0.5% moisture electrodes were obtained by baking for 8 hours at 250°F, and the 0.17% electrodes were obtained by baking for 1 hour at 825°F. Moisture content was determined in accordance with Mil-E-18038A (16 July 1957).

The results of these experiments performed on Grade M plate coated with 1100 aluminum alloy are given in Table I.

TABLE I

EFFECT OF COATING MOISTURE CONTENT ON E 7016 WELDS

Electrode	% Moisture	X-Ray Results	Longitudinal Bend Test Results
3/16 in. E 7016	1.2	some porosity	27°
3/16 in. E 7016	0.5	no porosity	25°
3/16 in. E 7016	0.17	no porosity	90°

It was concluded from this that low-hydrogen electrodes intended for use on aluminized steel should be kept dry for optimum results, although only slight porosity and no cracks were found at moisture levels of 1.2% by weight in the electrode coatings.

In addition, cruciform tests were made using the same plate and coating combination with 3/16 in. diameter E 7016 electrodes at the 1.2% and 0.17% moisture levels. No cracking was found.

2. Performance of Aluminized Steel When Subjected to Cruciform Test

The cruciform test was used in this investigation as a means of evaluating the crack susceptibility of welds made in aluminized plate. Single-pass welds were used in all cases. After welding, the specimens were sectioned and examined. No cracking was found. See Table II.

3. Mechanical Properties of Welds in Aluminized Steel

a. All-Weld-Metal Tensile Properties

A double-vee (120° included angle with 1/16 in. land) Grade M specimen coated on all surfaces, including the bevel, with 1100 aluminum alloy was welded using a 3/16 in. E 7016 electrode with 0.17% moisture in the coating. Type 12 (0.178 in. diameter) Hounsfield specimens were machined from the weld center as measured from top to bottom and side to side. All the tensile specimens from the 1100 alloy coated steel welds broke in a brittle manner.

In addition, two square-butt joints were prepared with a single bead deposited from one side only. One pair of Grade M test plates were coated with 1100 aluminum alloy prior to welding, while the other pair was coated with K726A alloy. The Hounsfield specimens taken from the K726A welds showed considerably more ductility than similar specimens obtained from 1100 alloy coated steel welds.

Representative specimens from the base metal and from welds made in uncoated steel with low-hydrogen electrodes were also tested for comparison. The results of all these tests are shown in Table III.

TABLE II

PERFORMANCE OF ALUMINIZED STEEL WELDS SUBJECTED TO CRUCIFORM TEST

Electrode	Manufacturer	Size, in.	Coating Alloy	Test Result
		(Base Metal--Grade M Steel)		
E 6010	B	3/16	1100	Porosity--no cracks
E 6010	B	1/4	1100	Porosity--no cracks
E 7016	B	5/32	K726A	Sound welds
E 7016	B	3/16	K726A	Sound welds
E 6016	C	5/32	K726A	Sound welds
E 7016	A	5/32	K726A	Sound welds
E 7016	B	3/16	1100	Sound welds

TABLE III

TENSILE PROPERTIES OF ALUMINIZED STEEL WELDS

(Base Plate - Grade M Steel)

Specimen No.	Coating Alloy	Joint Type	Yield Strength, psi	Ultimate Strength, psi	Elong-ation, %	Reduction of Area, %	Weld Analysis					
							C	Si	Mn	Al	P	Mo
1	1100	Fig. 3	-	92,000	3	3	.21	.57	.79	.72		
2	1100	Fig. 3	-	98,400	6	5	.21	.57	.79	.72		
3	1100	Fig. 3	-	83,000	3	1	.21	.57	.79	.72		
4	K726A	Sq. butt	86,,700	106,500	14	18	.21	.81	.76	.44		
5	K726A	Sq. butt	-	96,800	14	28	.21	.81	.76	.44		
6	1100	Sq. butt	-	86,000	4	3	.26	.23	.44	.39		
7*	none	none	33,400	58,800	34	55	.23	.02	.48	.02	.008	
8*	none	none	33,400	60,500	(a)	55	.23	.02	.48	.02	.008	
9(b)	none	Fig. 3	49,700	72,300	36	66	.16	.27	.61	.02	.026	.02
10(b)	none	Fig. 3	56,000	71,300	45	65	.16	.27	.61	.02	.026	.02

* Base plate specimens--not welds.

(a) Specimen twisted, and the elongation could not be measured. However, it appeared to be as ductile as specimen 7.

(b) The electrodes and base metal combination used for this comparison weld were the same as that used for all the uncoated physical property determination and explosion test specimens.

These results show that the 1100 aluminum increases the strength of the weld metal, but lowers its ductility to a dangerously low level. The K726A specimens showed considerably more ductility than was exhibited by the 1100 specimens. Micrographic examination and microhardness surveys revealed the difference to be the composition of the entrapped and partially dissolved coating. In the case of the welds in the 1100 alloy coated steel, a considerable amount of extremely hard aluminum compound, believed to be a mixture of Fe_2Al_5 and $FeAl_3$, was found in the root area and fusion line. These areas had microhardnesses of 500-600 Vickers. Similar appearing areas were found in the K726A alloy welds, but the hardness was only 250-300 Vickers (approximately the same as for the weld metal), indicating that the compounds found in the 1100 alloy welds had not formed.

b. Transverse Tensile Properties

A double-vee (120° included angle with a 1/16 in. land) butt joint using Grade M plate coated on all surfaces, including the bevel, with 1100 aluminum alloy (see Figure 3) was welded using 3/16 in. E 7016 electrodes with a coating moisture content of 0.17%. Subsize transverse tensile specimens (0.250 in. diameter) were machined from the weld and tested with the results as shown in Table IV.

All specimens necked on both sides of the weld, and the final fracture in all cases was in the base metal.

c. Transverse Bend Properties

A double-vee (120° included angle with 1/16 in. land) butt joint using Grade M plate coated on all surfaces, including the bevel, with 1100 aluminum alloy was welded using 3/16 in. E 7016 electrodes with a coating moisture content of 0.17%. Transverse bend specimens were machined and were bent around a 2 in. diameter mandrel. All the specimens bent without fracture, but all of the bending took place in the base metal adjacent to the welds.

d. Longitudinal Bend Properties

The longitudinal bend tests of welds made in 1100 coated plate indicated very low ductility. See Section C, Item 1.

TABLE IV

ALUMINIZED WELDMENT TRANSVERSE TENSILE PROPERTIES

Specimen No.	Yield Strength, psi	Ultimate Strength, psi	Reduction of Area, %	Failure Location
1	54,000	72,000	32	Base Metal
2	50,000	67,000	34	Base Metal
3	50,000	70,000	33	Base Metal
4	50,000	69,600	33	Base Metal

e. Longitudinal and Transverse Fillet Weld Shear Properties

Test specimen components of Grade M plate were prepared in accordance with the dimensions given on page 1460 and 1461 of the Welding Handbook, 3rd Edition. Representative specimen components were then coated with 1100 aluminum alloy and K726A, while the remainder were left uncoated.

These specimens were then welded using 3/16 in. diameter E 7016 electrodes. The results are presented in Table V.

f. Notch Toughness

The notch toughness of welds made in aluminized (1100) steel using the double-beveled weld specimens previously described, as measured by the Charpy Vee-Notch Test, is very poor at room temperature. The 10 ft-lb transition temperature is in the order of 200°F (see Figure 4).

As can be seen, the curve goes up sharply in the temperature range 200°-264°F. The path of failure for the high values was through base metal, fracture path A, Figure 4, even though the notch was located in weld metal. The lower values correspond to failure through weld metal, fracture path B.

It is not obvious why the path of failure should vary within this temperature range. It may be strictly a statistically arbitrary effect caused possibly by variations in the geometry of the notch root.

g. Performance of Fillet Welds in Aluminized Steel Plate Subjected to Explosive Loadings

A box-type specimen as shown in Figure 5 was developed for use in evaluating the relative performance of fillet welds in uncoated and aluminized Grade M steel when subjected to explosive loadings. The specimens were welded using 3/16 in. diameter E 7016 electrodes which had been baked for one hour at 825°F and stored at 250°F until used. All the specimens were manually welded.

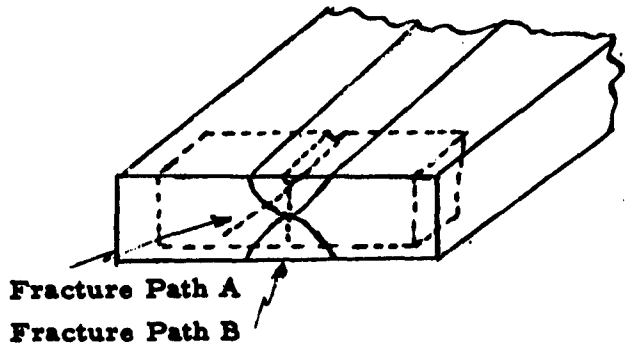
The experimental arrangement is shown in Figure 6 and consisted of an explosive charge down the center of the specimen. The charges were all 10 in. lengths of Primacord ranging from 40 to 400 grains of explosive per foot of length. By using various sizes together or separately, a whole range

TABLE V

LONGITUDINAL AND TRANSVERSE FILLET WELD SHEAR PROPERTIES

Specimen No.	Plate Coating	Shear Stress at Failure, psi	Location of Failure
	Longitudinal Shear Specimens		
1	none	(a)	Base Metal
2	none	(b)	Base Metal
3	1100	32,500	Weld Metal
4	1100	33,600	Weld Metal
5	K726A	40,600	Weld Metal
6	K726A	44,300	Weld Metal
	Transverse Shear Specimens		
1	none	79,400	Weld Metal
2	none	75,500	Weld Metal
3	none	82,500	Weld Metal
4	none	56,000[c]	Weld Metal
5	1100	43,000	Weld Metal
6	1100	46,700	Weld Metal
7	1100	39,400	Weld Metal
8	1100	35,400	Weld Metal
9	K726A	(d)	Base Metal
10	K726A	(e)	Base Metal
11	K726A	(f)	Base Metal
12	K726A	68,400	Weld Metal

(a) Weld metal stress at failure of base metal was 51,300 psi.
(b) Weld metal stress at failure of base metal was 54,700 psi.
(c) One weld of this specimen had excess porosity.
(d) Weld metal stress at failure of base metal was 74,300 psi; Specimen was retested and base metal failed again at a weld metal stress of 82,200 psi.
(e) Weld metal stress at failure of base metal was 81,500 psi.
(f) Weld metal stress at failure of base metal was 77,700 psi.



Location of Impact Specimens, Notch, and Path of Failure

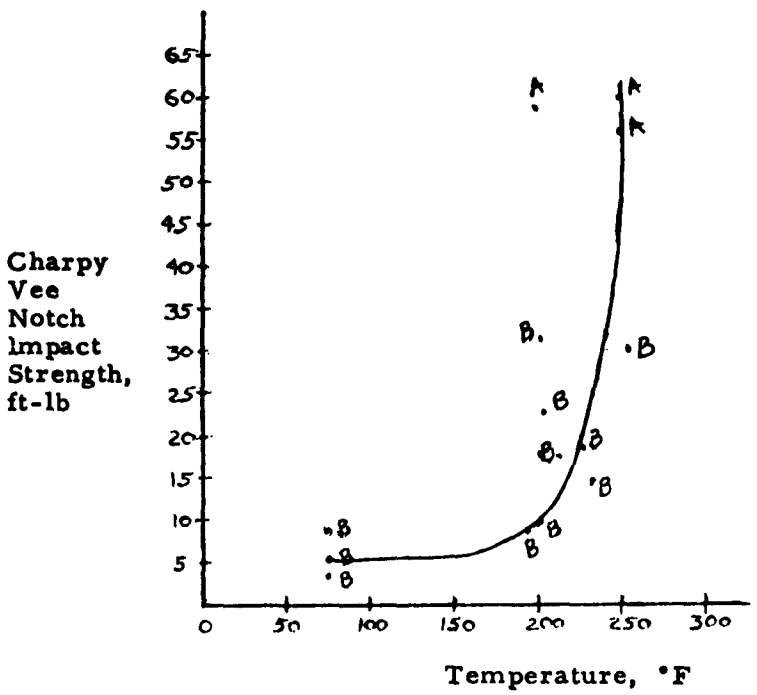


Figure 4 - Notch Toughness of E 7016 Welds Made in Grade M Steel Coated With 1100 Aluminum.

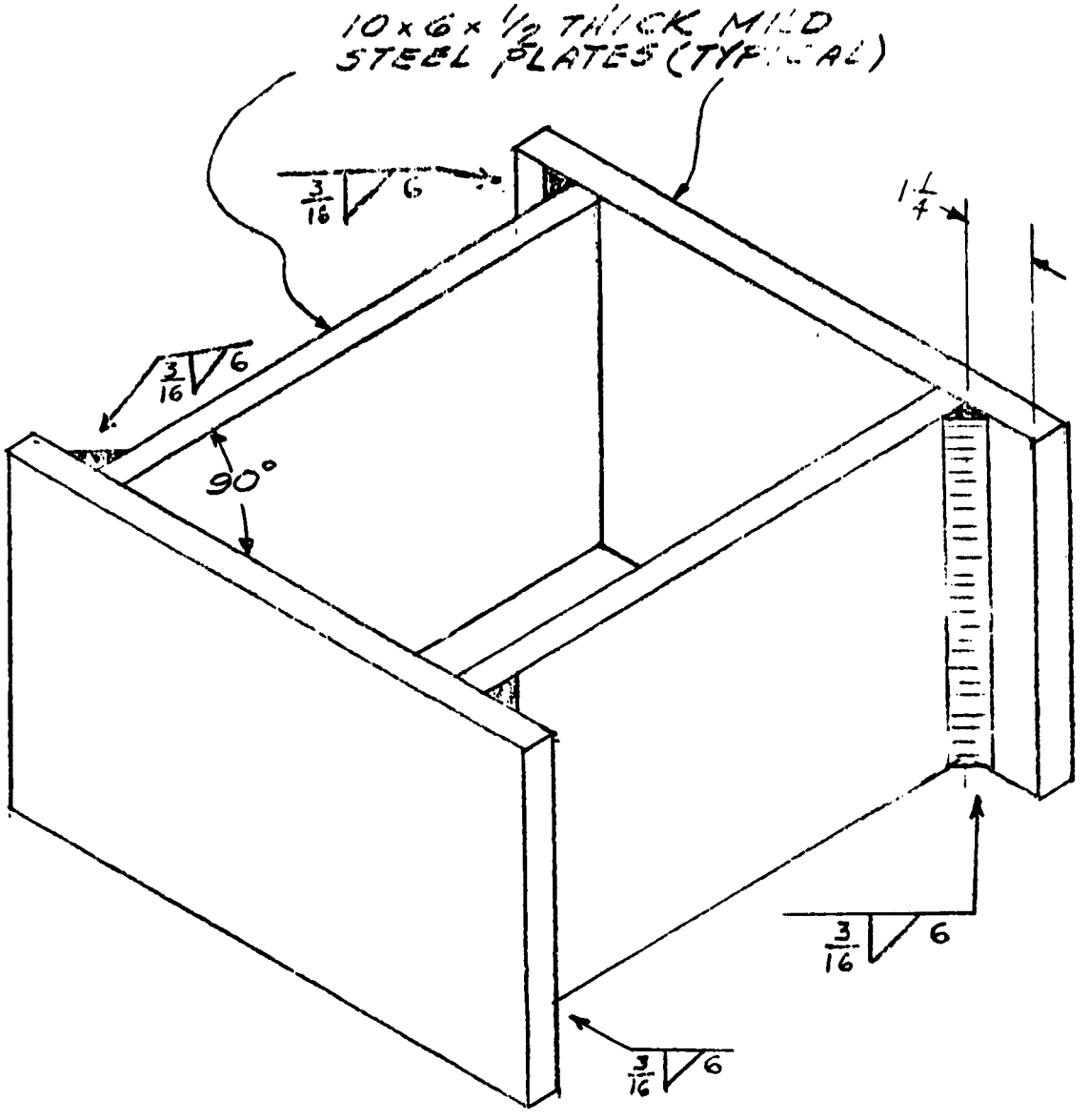
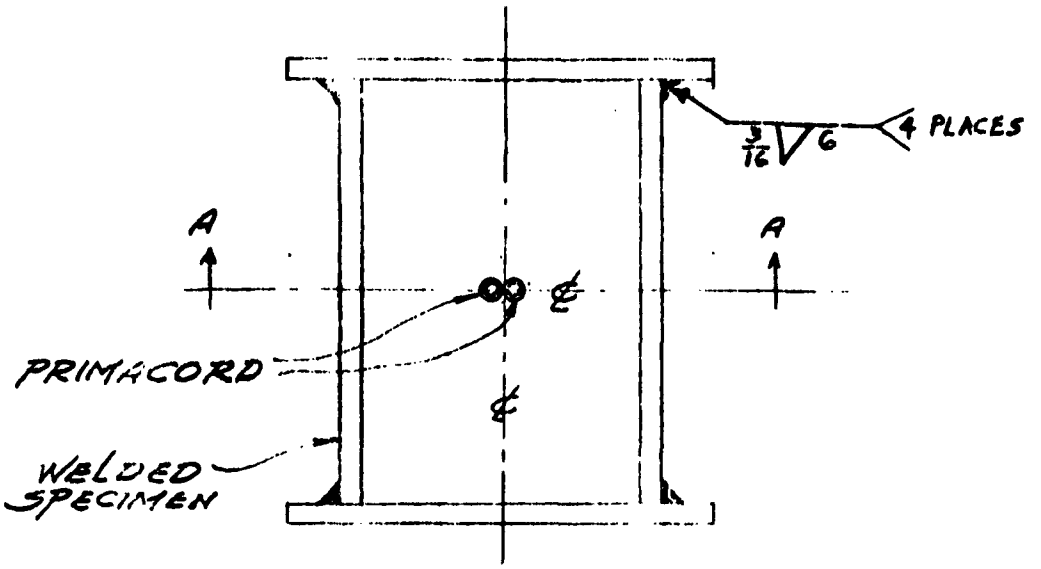
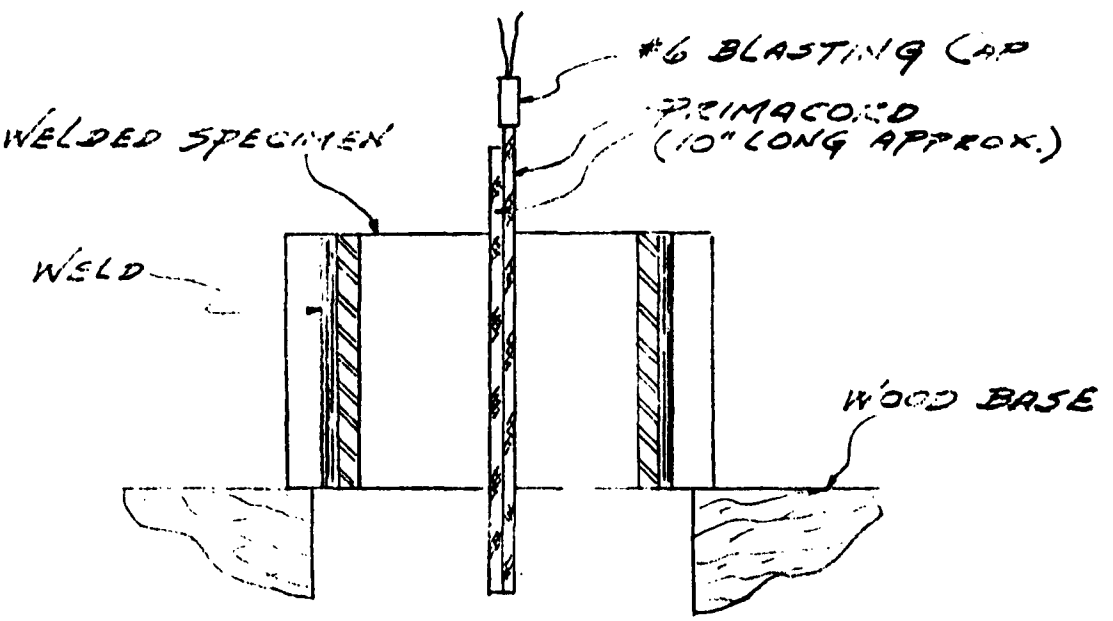


Fig. 5 EXPLOSION TEST SPECIMEN



TOP VIEW



SECTION A-A

FIG. 6 EXPERIMENTAL ARRANGEMENT

of core loadings was obtainable. The charge was initiated with a No. 6 blasting cap, and failure was arbitrarily defined as the fracture of more than one-half of one weld.

Test temperature ranged from 10-30°F and there was no apparent effect of temperature on these specimens within this range.

The results are depicted in Figure 7.

Additional tests were run on coated specimens welded with the Fer-Al* electrodes developed by one manufacturer for use on aluminized steel. The results were the same as obtained using regular E 7016 electrodes.

The results indicate the great superiority of the performance of uncoated plate specimens over 1100 aluminum alloy coated material. Typical failures are shown in Figures 8 and 9. The average uncoated specimen withstood approximately 4 times the explosive charge which caused failure of the coated specimens.

The range wherein some specimens passed and some failed is wider than necessary for the uncoated specimens because some of the specimens which passed a given charge were retested at another charge level. Generally, these specimens failed at a lower level than those specimens subjected to only one firing. It would have been desirable to have used a separate specimen for each firing since the effect of prior damage in some of the specimens subjected to more than one charge was unknown, but this was not practical within the limits of time and funds. If a separate specimen were used, the technique would have been to establish a "50% failure point" for each variable investigated, thus establishing a direct comparison. In the "50% failure point" method, the charge weight from one shot to the next is increased or decreased depending on whether the previous specimen failed or did not fail, and a fresh specimen is used each time.

IV. DISCUSSION OF RESULTS

As the previous paragraphs indicate, most of the work on this program was done using Grade M steel base metal coated with 1100 aluminum alloy. This combination was selected because it was considered to be the least

* Tradename

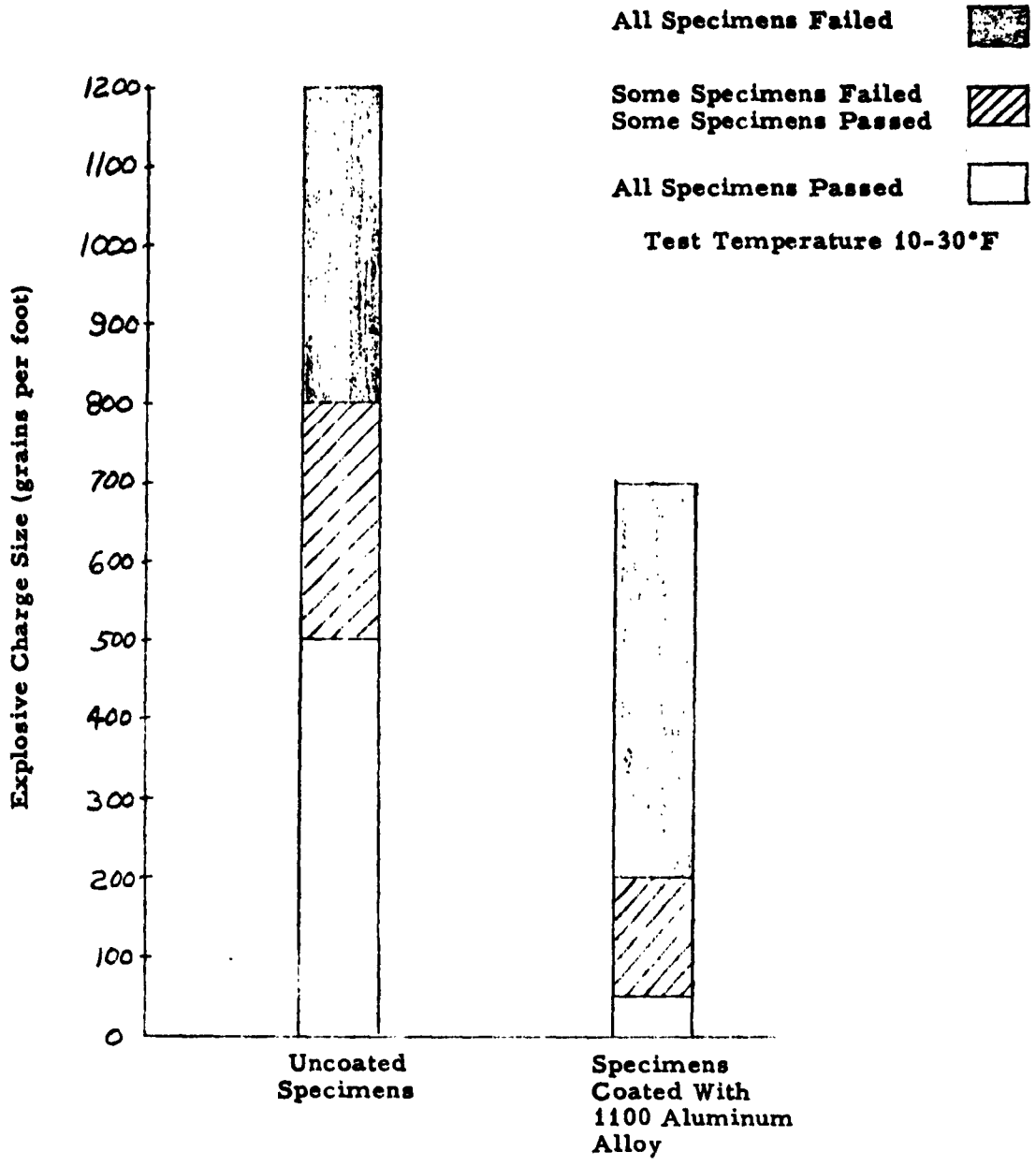


Figure 7 - Explosive Impact Performance Comparison

Figure 8 - Typical Failures of Uncoated Grade M Specimens Subjected to Explosive Loading

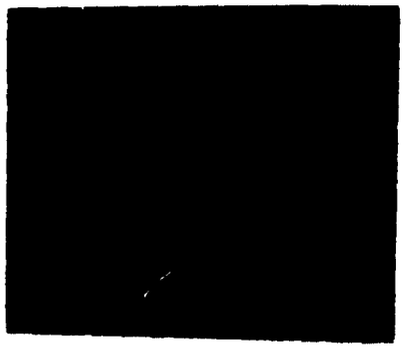
Figure 9 - Typical Failures of Grade M Specimens, Coated With 1100 Aluminum Alloy, when Subjected to Explosive Loading

critical combination and the one with the best possibility of having good weldability. The plan was to extend the work to Grade M steel coated with K726A alloy, and then to the HT Grade, first coated with 1100 alloy and then with K726 alloy after the weldability of the Grade M-1100 aluminum alloy combination was established.

The work shows that the performance of the welds in aluminized steel of the selected combination does not compare favorably with that of welds in uncoated steel. The basic difference in the two types of material is the aluminum coating which is melted and introduced into the weld. Since aluminum is soluble in iron up to 36%, all of the aluminum in the weld area can theoretically be taken into solution in the weld. However, as is shown in Figures 10, 11, and 12, this does not occur. Partially dissolved aluminum was found entrapped in the toe, fusion line, and root of the welds, and a series of unidentified phases, presumed to represent all the possibilities in the iron-aluminum alloy system, was found in the areas surrounding the entrapped aluminum. This finding is consistent with the fact that the time involved in producing a weld is quite short and that, as a result, mixing of molten base metal and weld metal is inhomogeneous.

It was expected that the curved fusion line plane which includes the entrapped aluminum would represent a plane of weakness and that the path of fracture of such welds when subjected to the explosion test would follow this plane. A metallographic examination of such a fractured weld showed that the fracture started in the aluminum compound in the root of the fillet weld (see Figure 13) and did follow the general path described by the fusion line, but did not always go through the aluminum-rich fusion line phases. The path of fracture in all cases examined was slightly on the weld side of the fusion line. However, since the cross-sections examined represent only a few of the infinite cross-sections possible and since all the entrapped aluminum and aluminum-rich phases do not lie in the same plane, it is still probable that the path of failure was through the plane containing the largest amount of these phases.

An attempt was made to determine the phases present in the fracture plane by scraping the surface of a fracture and using X-ray diffraction. The



Neg. No. 19946 50 X
2% Nital

Toe of Fillet Weld Showing Entrapped and Partially Dissolved Aluminum.

Figure 10

Neg. No. 19945R 50 X
2% Nital

Root Area of Fillet Weld Showing a Series of Fe-rich Al Alloys. The largest areas are Fe_2Al_5 and $FeAl_3$.

Figure 11

Neg. No. 19945 50 X
2% Nital

Root and Fusion line of Fillet Weld Showing Partially Dissolved Al and a Series of Al-rich Phases.

Figure 12

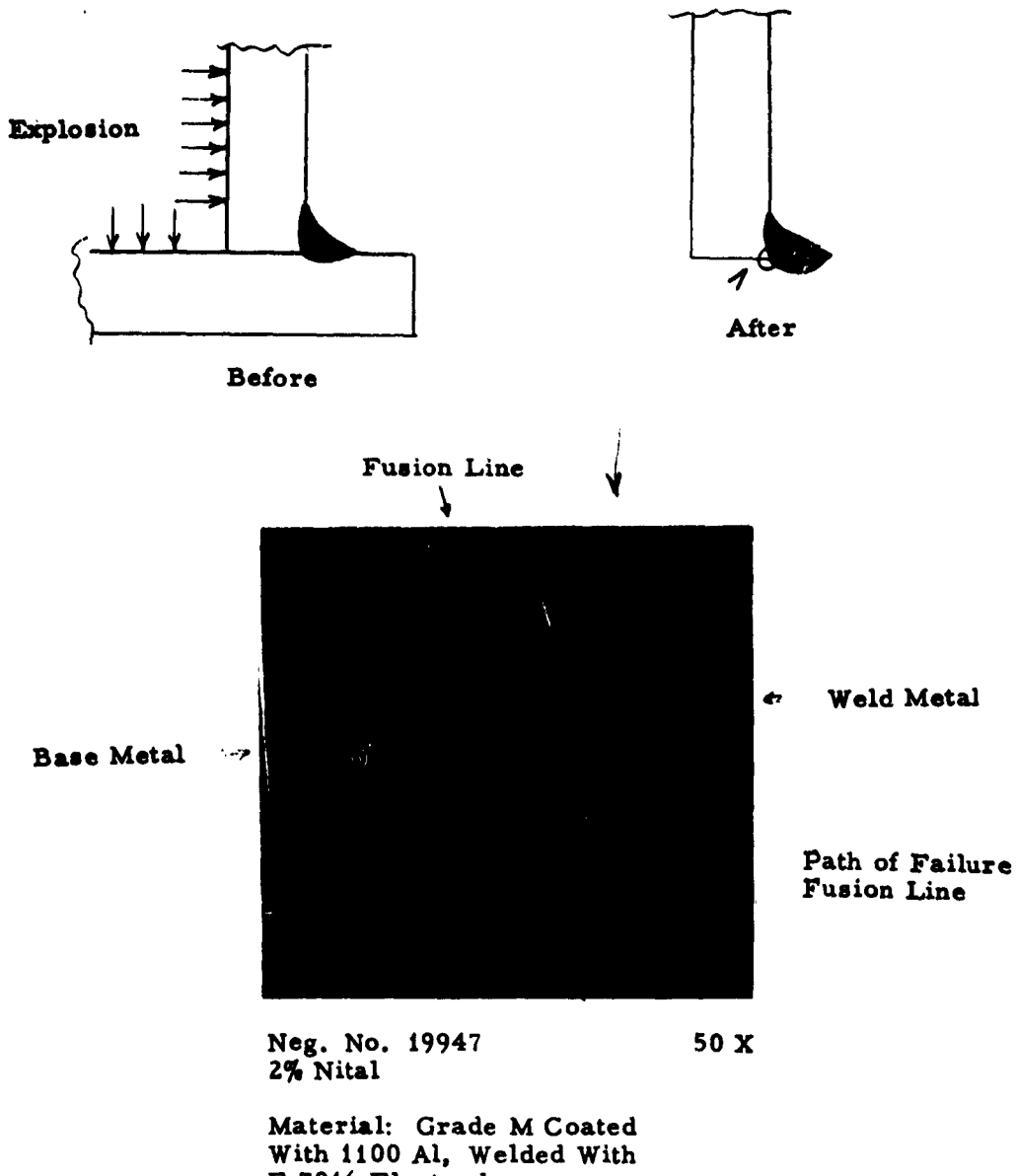


Figure 13 - Area of Failure Initiation in Root of Fillet Weld Exposed to Explosive Loading

only phase appearing was alpha iron, and the sample was magnetic indicating an aluminum content less than 16%. These results do not correlate with the microscopic observations, and hence it was concluded that the sampling technique was inadequate.

A microhardness measurement was made of some of the larger grains of the aluminum-rich phase shown in Figure 11, and it was found to have a hardness of 525 Vickers. This compared to a hardness of 572 for the Al-Fe compound between the steel and the aluminum coating, 192 Vickers for the base metal, 43 Vickers for the aluminum coating, and 270-290 Vickers for the upper regions of the weld metal.

Since the hardness of the Al-Fe compound between the aluminum coating and base metal was very nearly the same as for the compounds found in the welds, an X-ray diffraction pattern was obtained from some of this material and both Fe_2Al_5 and $FeAl_3$ were found in addition to aluminum and alpha iron. These aluminum-iron compounds Fe_2Al_5 and $FeAl_3$ are therefore considered responsible for the poor performance of the welds in the 1100 aluminum coated steel.

The impact test results were unfavorable to the aluminized steel. The double-vee butt joint was selected as the best type from which to obtain an impact specimen that might behave like the root area of a fillet weld since it was not possible to obtain samples directly from the root area of fillet welds. The impact specimens were cut from the weld with the longitudinal axis of the specimen parallel to the longitudinal axis of the weldment, and a vee notch was then cut transverse to the weld axis. (see Figure 4).

At all test temperatures, the weld fractured brittlely, and there were sizable areas showing intergranular failure indicating the presence of a low-melting brittle phase. At room temperature (80°F), the average value determined was 5 ft-lb. At temperatures from 200°F to 264°F, the results were scattered and depended on the path the fracture took, an apparently arbitrary effect.

All-weld-metal tensile specimens, equivalent to the areas from which the impact specimens were obtained, also broke in a brittle manner.

It was concluded from these findings that aluminized (1100) steel welded with conventional low-hydrogen electrodes is unsatisfactory for naval ship construction. It is also concluded that to weld aluminized steel for satisfactory performance, the aluminum (1100) coating will have to be removed either before welding by mechanical means, or during welding by fluxing, using a preplaced flux or a flux incorporated in, or superimposed on, the present electrode coatings.

Experiments to determine whether the aluminum could be fluxed from the weld area were without success. The first attempts were made using Fe_2O_3 preplaced, in an effort to produce the thermit reaction ahead of the arc:

$$2\,Al + Fe_2O_3 \longrightarrow Al_2O_3 + 2\,Fe$$

The reaction did not occur or, if it did, did not go to completion because aluminum was still found in the welds. The preplaced Fe_2O_3 in a sodium silicate carrier seemed to de-wet ahead of the arc rather than react with the aluminum.

A preplaced flux composed of 20% $CaCO_3$, 20% CaF_2, and 60% - 40 Baumé Na_2SiO_3 was more effective on bead-on-plate welds, but was ineffective on fillet welds.

Lastly, to get a quantitative result from experiments to oxidize the aluminum, E 7016 electrodes were dip coated in a flux composed of 8 1/2% anhydrous sodium silicate, 71% nickel oxide, and 20.5% water by weight. These electrodes were dried for 18 hours at 250°F, then baked for two hours at 800°F, and stored at 250°F until used.

Welds made with these electrodes analyzed at 5.55% nickel, showing the reduction of the nickel oxide; but the aluminum content of the weld was 0.61%, which showed that the reduction of the nickel oxide was primarily by the steel and not by the aluminum.

The program funds were exhausted at this point, and so the work was discontinued. The next step, however, was to be work with fluxes containing highly reactive fluorides and chlorides in an attempt to convert the aluminum to volatile aluminum chloride or fluoride. Such fluxes are a last resort because of the health hazard associated with their use.

Although the majority of the work on this program was done using Grade M plate coated with 1100 aluminum alloy, a limited amount of data was obtained from welds Grade M plate coated with K726A aluminum alloy. Specifically, the welds were studied for crack resistance using the cruciform test, all-weld-metal tensile properties were obtained, and welds were subjected to the longitudinal and transverse fillet weld shear tests. The results of these tests were more favorable than those obtained with 1100 aluminum alloy. The all-weld-metal tensile test revealed somewhat more ductility and higher strength, and no cracking was observed in the cruciform specimens.

A microscopic examination of some of these welds revealed the entrapment of aluminum-rich phases in the toe, fusion line, and root of the welds similar to the findings with the welds in the 1100 aluminum coated plates. However, two significant differences were also found. The interface between the K726A aluminum coating and the steel appeared to be free of iron-aluminum compounds, while the hardness of the entrapped aluminum-rich phases was about 300 Vickers, the same as the hardness of the weld metal itself, and also the same as the hardness of the K726A aluminum coating. It is speculated that the absence of the aluminum-iron compound region between the coating and the base metal may have resulted from the lower temperature and shorter times used during the coating process, but it is also possible that the addition of the silicon changed the wetting characteristics of the coating so that it was not necessary for the aluminum to alloy with the steel surface in order to obtain an adherent coating.

The apparent absence in the K726A coated steel welds of the Fe_2Al_5 and $FeAl_3$ compounds found in the welds of the 1100 coated steel is not so readily explained. It is possible that, in the case of the 1100 aluminum coated steel welds, the compounds found were not formed during welding but were formed during the coating process and were entrapped during welding. If this were true, then it could be assumed that the aluminum-rich phase found in the K726A coated steel welds is practically undiluted K726A alloy and that the Fe_2Al_5 and $FeAl_3$ compounds, while not identified by the hardness measurements, are actually present in small amounts surrounding the islands of K726A phase, since there must be a composition gradient established between the molten coating and the molten weld metal even in the short time involved

in making a weld. If this can be assumed,* then the results of the tests indicate that the small amount of Fe_2Al_5 and $FeAl_3$ formed as a result of the actual welding are not so detrimental as is the case when the compounds are preformed and introduced into the weld as a massive inclusion. In any event, the mechanical properties, microhardness, and microstructure of the welds made on steel coated with K726A offer encouragement that the solution to successful welding of aluminized steel lies in modification of the coating alloy. This is one obvious direction for future effort.

V. CONCLUSIONS

1. To obtain porosity-free welds in mild steel coated with aluminum or aluminum alloys where aluminum is introduced into the weld, it is necessary to use low-hydrogen electrodes.

2. Grade M steel coated with 1100 aluminum alloy and welded with E 7016 electrodes is not a satisfactory material for naval ship construction because of the poor performance of such welds when subjected to explosion impact loadings.

3. The main cause of the poor performance of fillet welds in Grade M plate coated with 1100 aluminum and welded with E 7016 electrodes is the entrapment and partial dissolution of aluminum and the presence of aluminum-iron compounds Fe_2Al_5 and $FeAl_3$ in the toe, fusion line, and root areas of the weld.

4. To obtain welds in steel plate coated with 1100 aluminum, comparable to welds in uncoated steel, it is necessary to remove the aluminum from the faying surfaces of the weld either by fluxing or by mechanical means.

5. Preliminary investigation of welds made in Grade M plate coated with K726A aluminum alloy produced much more encouraging results. While the results still were not comparable to results obtained from welds in uncoated plate, they were considerably better than those obtained with 1100 aluminum coated steel welds. This indicates that the composition of the alloy

* The absence of $FeAl_3$ and Fe_2Al_5 may be entirely a constitutional effect--the phases may not precipitate out of Fe-Al-Si melts whereas they do if the Si is absent.

coating can be changed to offset the difficulties obtained when welding on steel coated with unalloyed aluminum.

VI. RECOMMENDATIONS AND FUTURE WORK

1. It is recommended that a limited study be made to develop fluxes capable of removing the aluminum from the faying surface of the weld. This study should include fluxes containing highly reactive halides.

2. A limited number of explosion tests should be performed on specimens coated with K726A aluminum alloy since preliminary results of all weld metal tensile specimens showed an improvement in ductility compared to the 1100 aluminum coated specimens.

3. The possibility of adjusting the coating composition to improve the weldability of aluminized steel offers considerable promise and should be investigated thoroughly.

4. The explosion test described in this report should be evaluated further to prove its usefulness in differentiating the performance of fillet welds. The effects of the testing variables (weld size, specimen size, and testing temperature) on the test results and relative performance of welds should be established.

VII. LOGBOOKS AND CONTRIBUTING PERSONNEL

This work is recorded in ARF Logbooks Nos. C-9125, C-9437, C-1174, C-1136, and C-9500.

The personnel contributing to this work were:

Harry Schwartzbart - Welding Research Supervisor
David C. Brown - Research Metallurgist - Project Leader
Robert Maze - Welding Technician
John Dorcic - Foundry Technician
Thomas Wonder - Welding Technician

The explosion tests were performed by ARF's Explosive Research Section under the supervision of James L. Austing.

Respectfully submitted,

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